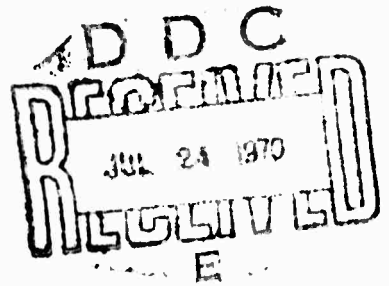


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IN
Ti-Mo ALLOYS

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BOEING COMMERCIAL AIRPLANE GROUP
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STRESS-INDUCED TRANSFORMATIONS IN TI-MO ALLOYS

M. J. Blackburn and J. A. Feeney

Several forms of martensite have been observed in titanium alloys after transformation of the high-temperature bcc form of titanium to hcp (1-9), orthorhombic (9,10), and fcc (8) or fct (5,6) structures. The fcc martensite is possibly related to spontaneous transformation (8,11,12), therefore is not characteristic of bulk material and will not be discussed here.

In the Ti-Mo system, it has been shown that hexagonal martensite is formed in alloys containing up to approximately 4% Mo and orthorhombic martensite is formed in alloys containing between 4% to 8% Mo (9). The M_s is below room temperature for alloys that contain greater than 10% Mo and thus remain single-phase bcc on water quenching. However, a "martensitic" structure can be induced by plastic deformation in alloys containing 10% to 17% Mo (2,12-14).

In pure titanium and in relatively dilute titanium alloys, the hexagonal martensite that is formed has a habit plane near $\{334\}_\beta$ (3,4,7). In more concentrated alloys such as Ti-(11-12)Mo and Ti-(5-6)Mn, habit planes near $\{344\}_\beta$ have been observed (1,2,8,13). In the Ti-Mn alloys the $\{344\}_\beta$ martensite was formed during quenching. The phenomenological theory of Bowles and MacKenzie (15) can adequately explain the characteristics of the $\{334\}_\beta$ -type bcc \rightarrow hcp transformation. Furthermore, a model has been

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proposed by Hammond and Kelly (8) to account for the $\{344\}_{\beta}$ habit plane in Ti-Mn alloys in which the product phase is also hexagonal. However, many characteristics of the $\{344\}_{\beta}$ -type martensite in Ti-Mo alloys remain unexplained. For example, it has never been conclusively shown that the structure of $\{344\}_{\beta}$ -type "martensite" in Ti-Mo alloys is hexagonal; indeed, one would expect an orthorhombic structure from the results of Bagariatskii et al. (9). Gaunt and Christian (13) had difficulty in reproducing the orientation relationship proposed by Liu (2) and in fact found only extra reflections on the bcc layer lines of a transformed single crystal of a Ti-12.5Mo alloy. Blackburn and Williams (12), using transmission electron microscopy, found that the "martensite" in a Ti-11.6Mo alloy had a bcc or bct structure and contained complex arrangements of internal twins. These authors were unable to determine the crystallographic characteristics of this stress-induced "transformation" because of the interference from a spontaneous transformation in their thin foils. A further unexplained feature is that $\{344\}_{\beta}$ -type "martensite" has not been observed to form in Ti-Mo alloys on quenching unless it is associated with $\{334\}_{\beta}$ -type martensite plates (2).

In an investigation of the microstructure and properties of the semicommercial alloy Ti-11.5Mo-6Zr-4.5Sn (often called Beta III), it was found that the bcc solid solution was retained on water quenching from temperatures greater than 840°C. Further cooling to liquid helium temperatures did not promote transformation to martensite. However, in the range -196° to +150°C, morphological features resembling martensite plates were formed during plastic deformation (Figs. 1a and 1b). It may be seen that the morphology changes with the amount of plastic strain: at low strains the platelike features are narrow and straight-sided for a considerable portion of their length

(Fig. 1a); at higher strains the plates thicken, assume a lenticular shape, and often contain several variants of internal striations (Fig. 1b). The volume fraction of plates increased with increasing strain from ~5% at 2% elongation to ~55% at 22% elongation. X-ray analysis of as-quenched specimens revealed the presence of bcc β phase and hexagonal ω phase. At plastic strains of 2% to 15%, no new phases were detected, but the ω reflections decreased in intensity. However, at the highest strain (22% elongation), a small amount of orthorhombic phase was detected with approximate lattice parameters of $a = 3.12 \text{ \AA}$, $b = 4.86 \text{ \AA}$, $c = 4.71 \text{ \AA}$.

The stress-induced "transformation" in Beta III was examined by thin foil electron microscopy. Fortunately, the spontaneous transformation that occurs in Beta III occurs to a lesser extent than in comparable Ti-Mo alloys. It will be shown that the platelike features formed at low strains are in fact mechanical twins. The absence of a stress-induced "product" with crystal structure different from that of the beta phase is therefore in agreement with the X-ray evidence. Figure 2a shows two such mechanical twins in a specimen deformed 2%. It can be seen that the twins relax on removal of the applied stress, leaving a highly damaged region A' adjacent to the straight interfaces of twin A. Note that twin B impinges on the original twin interface. Both twins contain a high dislocation density at this stage. Figure 2b illustrates the change in morphology at higher strains (6%). Note that the twins have thickened and that a second-order system is in operation. Where the internal twinning system intersects the primary twin boundary, displacements (or serrations) are produced. At the highest strains (15% to 22%) the complexity of the microstructure increases and three or more internal twinning systems may be observed.

Selected-area diffraction showed that the structure was bcc, thus establishing that the features were mechanical twins; however, the twinning system was not the $\{112\}$

$\langle 111 \rangle$ normally found in other bcc metals and alloys. Figure 3 illustrates these points; this and other diffraction evidence are consistent with a $\{332\} \langle 113 \rangle$ twinning system. Single-surface trace analysis (with all its attendant inaccuracies) was used to determine the habit plane of the twins in specimens deformed 2% to 5%. The results of 42 determinations are summarized in the standard triangle shown in Fig. 4.

In a future paper it will be shown that the phase transformations which occur in Beta III are very similar to those which occur in binary Ti-Mo alloys containing $\geq 11\%$ Mo (16). Therefore, two assumptions are made for the following discussion: (1) the stress-induced "transformation" is the same for Beta III and Ti-Mo alloys and (2) spontaneous transformation does not alter the crystal structure of the stress-induced product. The X-ray evidence indicates that the latter assumption is correct.

Details of the $\{332\} \langle 113 \rangle$ twinning system have been described by Richman (17) and by Crocker (18); the atomic movements involved are shown in Fig. 5, taken from Ref. 17. Note that in contrast to the simple shear of $\{112\}$ twinning, the $\{332\} \langle 113 \rangle$ mode involves shuffles as well as shears. Crystallographic information on the stress-induced "martensite" formed in Ti-Mo alloys is rather limited. It is of interest, however, to apply the above results to explain previous observations. First, the structure of the product phase is the same as that of the matrix, which explains the difficulty of previous workers (2,13,14) in establishing the orientation relationship for "hexagonal martensite." Second, the habit planes determined in this and other investigations is near $(344)_\beta$, but is also rather close to $(233)_\beta$. The deviation observed may be due to the second-order twinning that distorts the original habit plane, as

may be seen in Fig. 2b. Third, the shear direction measured by Gaunt and Christian (13) is near $\langle 113 \rangle$ and is in fact quoted as $\langle 338 \rangle$. Fourth, the magnitude of the shear has been measured (13) as ≈ 0.27 , which is in reasonably good agreement with the shear of $\{332\} \langle 113 \rangle$ twinning, which is $1/2\sqrt{2}$ or ≈ 0.30 . There is one discrepancy, however, between the observations on the secondary twin system in Ti-11.6Mo and that in Beta III: in Ti-11.6Mo the secondary twin system was shown to be $\{112\}_{\beta}$ (12), whereas in Beta III it is the same as the primary twin system, that is, $\{332\}_{\beta}$. It is not possible at present to account for this difference.

Finally, it is of interest to speculate why this rather obscure twinning system is observed in Ti-Mo alloys. There appear to be two possible explanations. First, as the molybdenum content is increased above 4%, the martensitic products change from hexagonal to orthorhombic (9); therefore the twinning may be related in some way to the orthorhombic transformation. If the results of Bagariatskii et al. (9) are extrapolated to 12% Mo, there is a tendency for the lattice parameters of the martensite product to approach those of the cubic parent phase. Second, the as-quenched structure is not single phase, but contains small particles of the ω phase. These particles are in general destroyed by the passage of twin interfaces during plastic deformation (16). If the atomic movements for $\{112\}_{\beta}$ and $\{332\}_{\beta}$ twinning modes are examined, it is found that the propagation of a $\{112\}_{\beta}$ twin interface through an ω -phase particle involves the noncooperative movement of atoms. More precisely, some planes of atoms have to move in opposite senses with respect to the shear direction. This situation is not true of $\{332\}_{\beta}$ twinning in which atomic movements in the ω phase are not dissimilar to those in the bcc phase.

Further work is in progress on a series of similar alloys of varying molybdenum content to establish whether

a relationship exists between orthorhombic martensite and the $\{332\}_\beta$ twinning mode.

ACKNOWLEDGMENTS

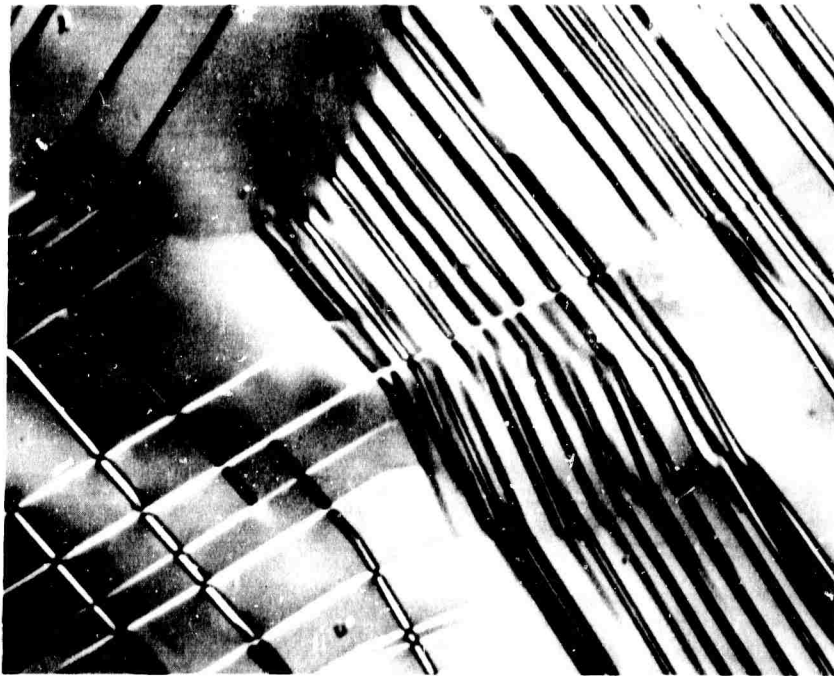
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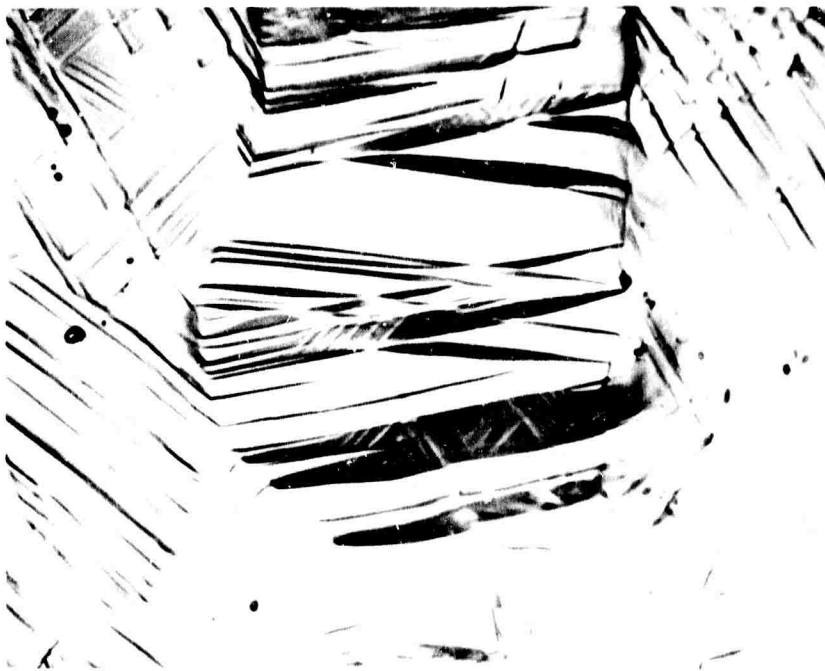
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(a) 2% uniform elongation.



(b) 10% uniform elongation

Figure 1 Optical micrographs showing twin morphology. (a) Normarski interference contrast, $\times 750$. (b) Bright field, $\times 750$.



(a) 2% uniform elongation.



(b) 6% uniform elongation.

Figure 2 Electron micrographs showing twin morphology and internal structure. (a) Note lateral relaxation of large twin from A' to A. Zone normal $\sim \langle 311 \rangle$. (b) Note that the primary twin boundaries are serrated by the second-order twinning system. Zone normal $\sim \langle 111 \rangle$.

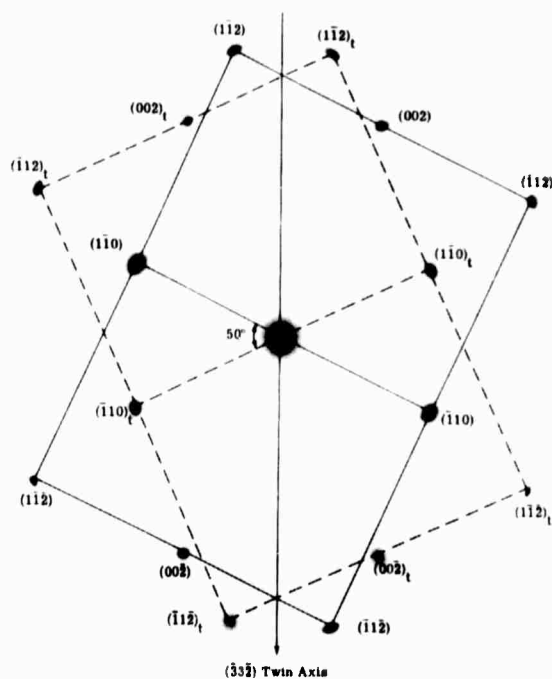


Figure 3 Selected-area diffraction pattern from a first-order twin. The twin and matrix $\langle 110 \rangle$ zones are related by 50° along their $\{002\}$ axes.

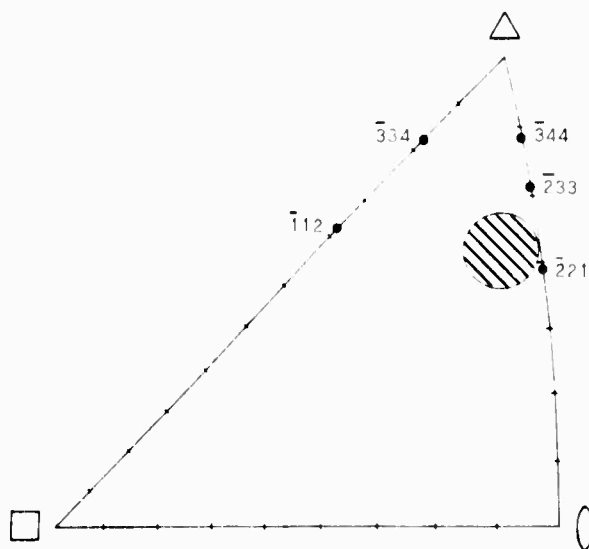


Figure 4 Stereographic plot which summarizes the habit plane determination of the primary twinning system. Only three of 45 determinations lay outside the shaded circle.

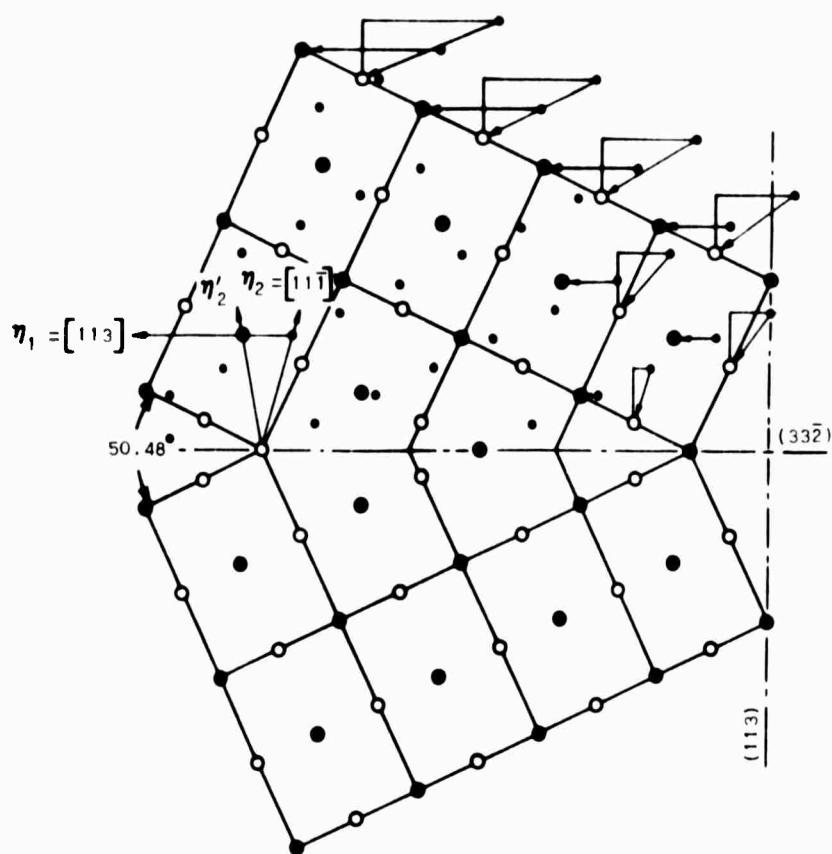


Figure 5 $(1\bar{1}0)$ projection of $(33\bar{2})[113]$ twinning in the bcc lattice. Note the 50.48° relation between the twin and matrix projections. Solid symbols represent atoms in the projection plane; open symbols represent atoms $0.707a_0$ above or below the projection plane (from Ref. 17).

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13. ABSTRACT The deformation characteristics of the metastable beta alloy Ti-11.5Mo-6Zr-4.5Sn (Beta III) have been studied. In the range -196 ⁰ to +150 ⁰ C, Beta III deforms primarily by mechanical twinning and to a lesser extent by slip and stress-induced orthorhombic martensite formation. The twin system is {332} <113> and not the {112} <111> system normally operating in body centered cubic materials.		

14. KEY WORDS	LINK A		LINK B		LINK C	
	ROLE	WT	ROLE	WT	ROLE	WT
Metastable Beta Titanium Alloy Deformation Characteristics Mechanical Twinning Stress-Induced Martensite						